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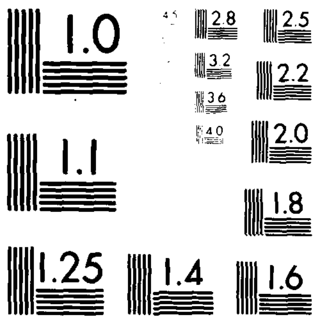

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20. Abstract

This review is an attempt to summarize the state of research and development in Europe into powder metallurgy of superalloys, and to ascertain which areas of the technology still require attention before the promised advantages of the powder metallurgy route are achieved.

Although 'superalloys' are normally taken to be those alloys whose bases are nickel or cobalt, the present review deals with materials which are currently in prospect of production by powder metallurgy and which find use in the hot parts of aeroturbines. Thus it includes an account of some ferritic materials and some titanium alloys. The review deals with disc and blade alloys and discusses powder production, consolidation and the relationships between processing, structure and the final properties of powder alloys.

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Difficulties associated with the casting and forging of very strong superalloys for aeroengine parts led, in the late 1960's, to an investigation of the powder route method of manufacture. It soon became evident that powder metallurgy offered a number of possible advantages compared to conventional procedures.

The very good chemical homogeniety of pre-alloyed powder metallurgy components as compared to cast structures offered major advantages in subsequent fabrication and in the integrity of the final product. For large forged discs this improved integrity was of considerable importance since uniformity of mechanical properties in several directions in the disc and reproducibility of properties (particularly low cycle fatigue strength) from disc to disc were often poor in conventional forgings. It was felt that the micro-scale homogeniety of powder components would lead to less difficulty in forging and that the forging problem would be further alleviated by the inherently fine grain size of the material.

The powder route also gave the possibility of introducing much larger volume fractions of hardening phases into alloys than could be conveniently achieved by casting. Some of these hardening components could be insoluable oxides which would give improved creep properties at the highest temperatures where γ' hardening was no longer a possibility.

Apart from the technical advantages, the economics of the powder metallurgy route began to look increasingly attractive as material costs increased, availability became more difficult and labour costs rose significantly. The cast/forge/machine route to superalloy components was extremely wasteful in material (some discs having a material utilization of less than 5%) and required heavy machining. If components could be formed from powder directly into near net shape, considerable savings might be possible.

These possible advantages led to intense activity during the 1970's into powder production methods, methods of consolidation and finishing and the properties of powder metallurgy parts. An AGARD meeting in Ottawa in 1976 was generally very optimistic about the possibility of components, particularly discs, being in service in the very near future but in fact very few examples of parts in service have been reported to the present time. This review is an attempt to summarize the state of research and development in Europe and to ascertain which areas of the technology still require attention before the promised advantages of the powder metallurgy route are achieved.

Although 'superalloys' are normally taken to be those alloys whose bases are nickel or cobalt, the present review will deal with materials which are currently in prospect of production by powder metallurgy and which find use in the hot parts of aeroturbines. Thus it will include an account of some ferritic materials and some titanium alloys. The review will deal with disc and blade alloys and will discuss powder production, consolidation and the relationships between processing, structure and the final properties of powder alloys.

2. Powder Production

A wide variety of processes have been used in Europe for the production of powders suitable for eventual incorporation into aeroengine parts. Different approaches have been used for alloys of titanium than for those which are basically nickel or cobalt and the present situation with respect to plant for these materials is quite different. Nickel base superalloys can be produced as powders with little remaining technical difficulty and in quantities large enough for commercial use. Alloys containing large quantities of titanium can only pe produced as powders with equipment which is still at the pilot plant stage and considerable technical problems still have to be solves. Table 1 is a summary of the powder materials that have been produced together with a brief description of the powder properties and the location of the relevant plant.

2.1 Nickel and Cobalt Base Alloys

Inert gas atomization has been most widely used for the production of nickel based superalloys[1,2,3,4] and those alloys of cobalt which are low in titanium[5]. A major argon atomizing powder production and consolidation plant has been set up by Henry Wiggin and Co. Ltd. in the United Kingdom[1,2,6]. The equipment consists of a vacuum induction melting unit mounted over a 6 m atomizing tower. The melting unit can deal with 500 kg melts and operates at 10^{-3} torr. The furnace is equipped with various small additions feeders as well as sampling and control devices. It operates with selected raw materials or master alloys and practices have been established which allow for scrap recycling.

Prior to atomization the melting chamber is back filled with argon

TABLE 1 POWDER PROPERTIES

ALLOY †	PROC-ESS	REF	ORGANIZATION ♦	MELT SIZE	STARTING MATERIAL	UTILIZ-ATION	SIZE RANGE μm	MEAN SIZE μm	DENDRITE SPACING μm	TAP DENSITY %	GAS ANALYSIS p.p.m. O	N	H	A
APK 1	A.A	1,2	WIGGIN	500 kg	VAC. MELT	70 %	5-500	-	5	65	32	16	-	1·5
IN 100	A.A.	3,4	CREUSOT-LOIRE	-	VAC. MELT	-	20-120	50	-	-	100	-	-	-
IN 100	CLET	3,4	CREUSOT-LOIRE	-	VAC. MELT	100%	200-600	450	-	-	60	-	-	-
U.700	CLET	4	CREUSOT-LOIRE	-	VAC. MELT	100%	200-600	400	-	-	50	-	-	-
VO/VS *	N.A.	5	C.R.M	-	-	-	400-500	450	4	-	400	-	-	-
V13/V14 *	REP	5	C.R.M.	-	-	-	400-500	450	4	-	80	25	-	-
Ti 6Al 4V	EBRE	13	C.E.N.G.	2 kg	CAST ROD	85 %	100-1000	350	10	60	1200	20	12	-
Ti 6Al 4V	EBRD	14,15	LEYBOLD-HER.	60 Kg	CAST ROD	70/80 %	250-600	320	20	60	1350	40	5	-
Ti 6Al 4V	CSC	10,11	HARWELL	12 kg	CAST ROD	-	50-1000	350	-	65	1650	-	-	-
Ti 6Al 5Zr-	EBRE	13	C.E.N.G.	2 kg	CAST ROD	82 %	100-1000	500	10	65	1500	<100	6	-

† SEE APPENDIX 1
♦ SEE APPENDIX 2
* EXPERIMENTAL ALLOY

A.A. - ARGON ATOMIZATION
N.A. - NITROGEN ATOMIZATION
CLET - MODIFIED ROTATING ELECTRODE
REP - ROTATING ELECTRODE

EBRE - ELECTRON BEAM ROTATING ELECTRODE
EBRD - ELECTRON BEAM ROTATING DISC.
CSC - CENTRIFUGAL SHOT CASTING

and, after suitable slag skimming, atomization occurs through a ceramic nozzle. Powder falls through the atomizing tower and is collected in a water cooled hopper. The atomizer has been designed to prevent mishapen particles which would be produced by impaction with the atomizer wall. The powders are sieved and blended in argon prior to a final degassing and bagging for subsequent HIP treatment.

The properties of a typical powder metallurgy disc alloy (APK 1) produced by this method are shown in Table 1. There is a considerable range of spherical powder sizes in the product and the useful output has to be restricted to eliminate large particles which sometimes show entrapped argon porosity. Provided this restriction is adhered to, oxygen,nitrogen and argon levels are generally less than 32 ppm, 16 ppm and 1.5 ppm respectively and the powder has a tap density of about 65%. No estimates of the solidification rates have been published but an examination of particle micrographs indicates that the dendrite spacing is about 5 μm. Wiggins[1] and others[7] estimate that oxygen and nitrogen levels should be below 100 ppm and 50 ppm respectively so that material well below these limits is readily available. There is some possibility of tramp element pick up from the melting and atomizing refractories but this appears to have been controlled to within acceptable limits. It is clear that the argon atomization of nickel base superalloys is a well established process with no serious technical limitations and that there is considerable production capability available within the United Kingdom.

Some nickel base alloy powders have been produced by the CLET process [3,4,8]. This is a modification of Rotating Electrode Process in which a cast alloy stick is melted by means of a reciprocating heat source and subsequently rotated to produce the necessary atomization. The reciprocating source spreads the heating over the electrode end, reducing thermal stresses and allowing large diameters (up to 160 mm diameter) to be melted. There is no gas porosity in the particles so that no particle fractioning is required, leading to high yields. It is claimed that the process very readily takes into solution large carbides[9] and that this allows considerable eventual grain growth. This may be important for alloys where creep resistance is eventually a critical parameter.

Some work has also been done with the EBRD process. This method has mainly been used for titanium alloys and will be described in detail below. For nickel alloys it appears to be capable of producing powders with similar shapes and sizes to argon atomizing but with somewhat higher oxygen contents (60 ppm).

In view of the ability of the Wiggin argon atomizing plant to deal with scrap, its high production rate, its proved technical capability and its commercial availability, it is clear that argon atomizing will be the only commercially viable process for production nickel-base superalloy powders for some time to come.

2.2 Titanium Alloys

In comparison with nickel superalloys, the reactivity of titanium with interstitial solute elements such as oxygen, nitrogen and hydrogen and with refractory materials makes for considerable problems in devising suitable atomizing procedures. Inert gas atomization gives material with unacceptably high gas contents and the processes used in Europe all rely on ingot melting with ancilliary centrifugal action to break up molten films into suitably sized particles. Several different procedures are being evaluated.

Centrifugal Shot Casting (C.S.C.) has been under investigation at

Harwell (United Kingdom) for some years[8,9]. The process involves a stationary electrode of the material to be converted into powder and a rotating water cooled crucible. Heating is accomplished by means of an electric arc struck between the crucible and the electrode which causes the end of the electrode to melt and fall as drops into the crucible. Under the action of centrifugal force the melt moves up the crucible walls to the lip where it breaks up into droplets. The whole process of melting, atomization and solidification takes place under an argon atmosphere at pressures in the range 0.3 to 1.0 atmospheres. The particle shape is essentially spheroidal with a high packing density and purity can be readily maintained at the same level as the electrode. Cooling rates in the titanium droplets have been calculated at 10^5 °/s but no details of dendrite spacings are available. Cooling rates about one order of magnitude better than this may be obtained by using helium atmospheres. Recent modifications have included two consumable electrodes, one of which is rotated[12].

The Electron Beam Rotating Electrode process (EBRE) has been developed by C.E.N.G., France[13]. An ingot mould is mounted vertically and electrodes of 50 mm diameter are charged to it. The top end of the electrode is heated by an electron beam and the resulting liquid alloy is converted to liquid spherules by the rotation of the electrode at about 4000 revs/minute. The liquid particles partially solidify in flight and eventually impinge on a water cooled adjustable screen. The particles are maintained spherical by an adjustment of the relative surface tensions between the particles and the screen by the presence of a commercial ablation agent. The whole process operates in a vacuum of better than 10^{-4} torr and the 50 mm diameter ingot can be melted at a rate of approximately 50 mm per minute. About 85% of the ingot is atomized to spherical powder with about 9% of ingot material and about 1% of agglomerated powder recycled. No cooling rates are quoted, but examination of micrographs indicates a dendrite size of about 8 μm for titanium alloys.

The Electron Beam Rotating Disc porcess (EBRD) has been developed by Leybold-Heraeus GmbH[14,15]. The base stock material is a stationary electrode of diameter between 60 and 150 mm. The tip of this ingot is melted by a 90 kw electron beam and the molten alloy drips into a water cooled copper crucible rotating at between 1400 and 1500 r.p.m. Under the action of centrifugal forces, the molten metal moves towards the lip of the crucible. In order to carefully control this movement a second 30 kw electron gun plays on the crucible pool. Finally the metal is atomized at the crucible edge, this atomization being aided by a third electron gun of 30 kw power. The ejected spherules are deflected by a screen and solidified in flight to produce the alloy powder. The whole process proceeds under a vacuum of 10^{-4} torr and the grain size in spherical powder is about 20 μm.

Each of the processes outlined has advantages and disadvantages. Both of the electron beam melting procedures are forced to operate in vacuum. This leads to some difficulty in achieving sufficiently rapid cooling rates since heat loss during trajectory must be by radiation only. This is a considerable problem for titanium alloys which have a high latent heat of solidification (500 j/gm). Generally long flight paths will be required and in the case of ERBD processing sufficiently long trajectories are not available to produce 100% spherical powder. An appreciable fraction of splatted powder, produced by liquid impingement on the container walls, is usually obtained and such powder generally has a poor tap density. EBRE claims to have overcome this problem by using a preparatory ablative coating on a water cooled deflector screen. However, the coating itself may be a cause of carbon pick up, but this is at present not considered to be a major problem. CSC uses an argon melting atmosphere and hence convective cooling is available to aid rapid solidification with a consequent increase in spherical powder yield and a reduction in chamber size.

ERBE uses a rotating consumable electrode and this places a quite stringent requirement on electrode shape and integrity as compared to the non rotating electrode processes. CSC yields a powder which has essentially the same composition as the initial electrode, but the two electron beam melting processes give the possibility, at least in theory, of some refinement of light element impurities. Such refinement is certainly present in EBRD where quite unacceptable losses of light elements such as aluminium can occur. This loss is presumably associated with the relatively long exposure of the liquid metal to vacuum conditions. This difficulty appears to have been overcome by ensuring that melting rates do not fall below certain critical levels. At present, the above processes are batch type and as such the production rates are limited by the relatively small electrode sizes. However, Leybold-Heraeus have proposed a more continuous type of plant[15], which may allow continuous feeding and the use of scrap material with some metal refining ability. It is not clear whether any experimental work has been carried out on this proposal.

In general, the dendrite spacings of titanium alloys produced by centrifugal means is greater than that for inert gas atomized nickel powders and this means that compositional inhomogenieties will be somewhat greater. This is particularly true of those processes where cooling rates are lowest (i.e. electron beam melting processes).

Of the processes used for titanium alloy powder production, no one process seems to have a clear edge over any other. One of the difficulties of assessing the merits of the methods is that there is no very clear idea of the effects which various imperfections may have on the mechanical properties of final components. It is, for instance, not absolutely clear whether impurities deriving from arc melting have a deleterious effect on fatigue properties or not. Until matters of this sort are investigated properly, there will be little chance of establishing the best process for titanium alloy powder production, and all the processes are likely to remain at the experimental or pilot plant level.

3. Powder Handling

The requirement of maintaining oxygen and nitrogen levels in finished superalloy products below well defined limits[7] imposes certain restraints on powder handling. The powders described in Table 1 have a high surface area to volume ratio and this can result in high quantities of absorbed gases if they are exposed to air. The experimental evidence on any reduction in properties due to this effect is somewhat contradictory. For nickel base alloys little affect of air handling has been observed in some cases[16] but in others there has been a reduction in the levels of final mechanical properties[17,18]. A further difficulty with argon atomized powder is the entrappment of argon which may give rise to thermally induced porosity during subsequent hot consolidation. As a result of this possibility most large scale European operations ensure that powder handling is conducted in inert atmosphere followed by a thorough vacuum degassing at temperatures up to 500°C[2]. In some cases, the level of argon present in consolidated parts is used as a quality control check[1] before material is passed on for further treatment.

There is less certainty about handling procedures for titanium alloys. Kenaith[19] investigated air handling, nitrogen handling with vacuum degassing at 450°C and container filling under vacuum at the powder plant stage. Inert atmosphere handling was found to improve final mechanical properties. Other workers have found little difficulty with air handling[13]. Some further work is required to assess the extent of the necessity for the relatively expensive inert gas handling.

4. Property Requirements

The properties required in finished powder metallurgy parts varies with the application in the aeroengine. Turbine blades, which operate at relatively high temperatures and moderate stresses, require good creep strength and ductility as well as good high temperature corrosion resistance. On the other hand discs have to operate at only moderate temperatures and the major concern is with high static strength and good resistance to low cycle fatigue. In addition high cycle fatigue and thermal fatigue resistance is required. The actual mechanical properties achieved will depend in a complex way on the alloy chemistry and microstructure. This last will in turn depend on the detailed mechanical and thermal history of the material.

The strength, σ, of materials at high temperatures is a sensitive function of temperature, T, and the rate at which they are required to deform, $\dot{\varepsilon}$. In addition, the grain size is of importance and these quantities are generally thought to be related by

$$\sigma = A \sigma^{n} d^{m} \exp\left({}^{-Q}/RT\right) \quad \ldots\ldots\ldots\ldots\ldots\ldots \quad 4.1$$

where A, n, m and Q are material and structure constants and d is the grain diameter[20]. Of critical importance is the value of m. At high temperatures m is negative and hence the presence of many grain boundaries has a deleterious effect on creep resistance and creep ductility. On the other hand at low temperatures m is positive so that grain boundary strengthening occurs. The exact transition temperature is often unknown but in general disc materials will require small grain sizes and blades large grain sizes. The small grain sizes in discs will also improve low cycle fatigue strength. These differences in grain size require different approaches to the production of powder metallurgy alloys and the two applications will be treated separately.

5. Nickel based Disc Materials

A wide variety of routes has been followed in consolidating powder superalloys for turbine disc applications. The spherical shape of powders and their extreme susceptibility to interstitial solute gases rule out the normal processes of cold compaction and sintering. Instead, canning followed by a high temperature deformation process is required. The deformation processes available include conventional forging, isothermal forging, high temperature isostatic pressing (HIP), rapid high temperature isostatic pressing (RHIP) and high temperature extrusion. These various processes can be combined in several ways as shown schematically in Fig. 1 . The choice of route depends on its ability to achieve the necessary structure and properties, its ability to allow a relatively straight forward production schedule as compared with cast and wrought material and whether the route has been designed to produce significant savings in materials and machining costs.

A general review of research in disc production indicates that the most commonly accepted procedure is HIP. Conventional forging is generally easier than for cast materials since the chemical homogeniety gives greater ductility in bilets and the fine grain size associated with powder metallurgy products can make them superplastic with a resulting decrease in flow stress. A more advanced process in use is isothermal forging. This is essentially a slow forming operation with hot dies. The flow stress of the materials are considerably reduced by the reduction in strain rate as compared to conventional forging, deformation is not affected by chilled layers at the die interfaces and much greater strains can usually be achieved without cracking.

Several alloys have been used for research into powder metallurgy disc

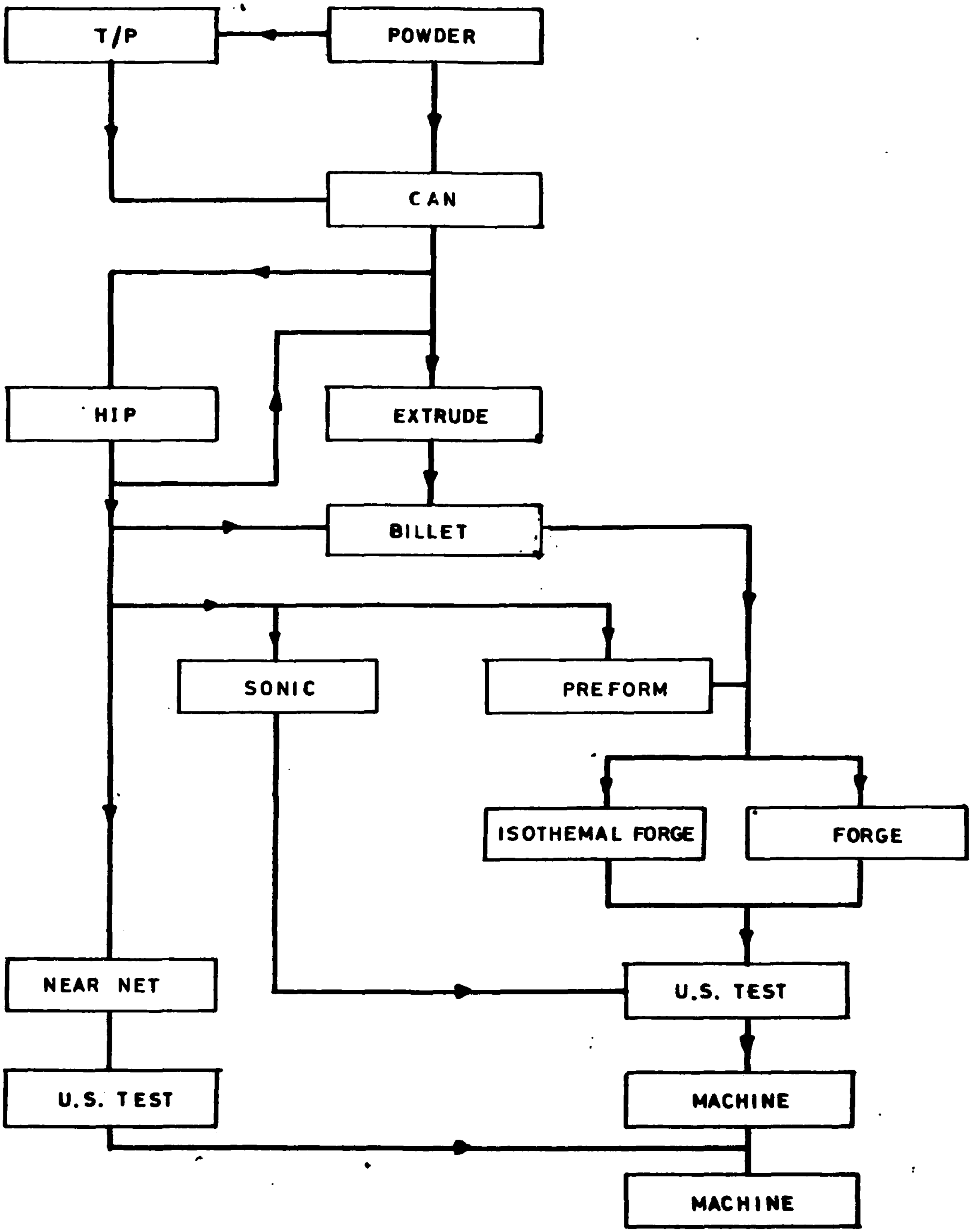


FIG 1 POWDER DISC PRODUCTION ROUTES

production. Some of these alloys were originally designed for use as blade alloys (e.g. IN 100) and were originally required to be cast since chemical inhomogeniety and strength prevented forging for use as disc material. The powder route offered a means of eliminating seggregation and also of reducing forging stress due to the fine grain size and considerable work has gone into the use of IN 100 as disc material.

Other alloys (such as ASTROLOY) were originally designed for disc use and the powder route makes their forging a simpler operation. Although a considerable amount of work is going on on disc materials, the large number of process routes available and the choice of alloys available makes comparision between various workers rather difficult. The various alloys will be discussed in turn.

5.1 Canning

Whichever processing route is chosen, material requires to be canned prior to consolidation. A number of different canning materials have been used and the shape of the can depends on the process route chosen. Simple consolidation to billet requires only simple shapes but if preforms or net shape HIP products are required the can shape must be more complex. Mild steel cans have been used most frequently[1,21,22,23] and forging preforms have been made in spun cans[1]. Stainless steel cans have also been used [12,24,25] but these are more expensive and are sometimes difficult to remove. A more complex process is being used in Germany[19,26] known as Electro Forming. Nickel is electroplated onto a wax preform which is subsequently melted out to give a suitably shaped container for HIP. Superplastic bag forming is also being investigated[27], but few details are available. Glass cans have also been used for extrusion procedures[28].

5.2 A.P.K.1 Powders

APK1 is a low carbon Ni-Cr-Co alloy specifically designed for powder metallurgy disc applications[1,2,27,29]. Initial work with existing powder alloys indicated that some deleterious structures could be obtained by the precipitation of carbides on prior particle boundaries. In APK1 this has been overcome by a reduction in carbon content and by careful control of HIP and heat treatment procedures.

The material has been processed by a number of routes including HIP and anneal, HIP and conventional forge and HIP and isothermal forge. In addition, a special procedure shown as thermoplastic processing (T/P) has been used. In this case powder is cold worked prior to consolidation either by ball milling or in a rolling mill. Care has to be taken to avoid oxygen pick up and from this point of view rolling mill operation is the most satisfactory. The object of the procedure is to allow recrystallization to a very small grain size ($\sim$ 1.5 μm) during consolidation by HIP. The resulting material is then very easy to forge, both conventionally and isothermally. The 650° tensile results and, where available, the low cycle fatigue properties are shown in Table 2 for normal powder and Table 3 for T/P processed powder.

The properties of HIPed and conventionally forged powder are dependent on whether the structure is finally 'hot worked' or 'warm worked'. In the warm worked condition the material shows a normal grain size with the grain boundaries replaced by a fine network of small dynamically recrystallized grains. These grains, since they are formed during the hot forging deformation, show a relatively high dislocation density which is retained through subsequent heat treatment. The structure is known as a 'necklace structure' and is thought to be

TABLE 2 PROPERTIES OF APK 1

H.I.P.	CONVENTIONAL	ISOTHERMAL	650°C TENSILE			CREEP 760 MNm^{-2}, 705°C		L.C.F. CYCLES
TREATMENT	FORGE	FORGE	·2% P.S.(MNm^{-2})	U.T.S.(MN/m^{2})	ELONG %	LIFE (HRS)	ELONG. %	0-1080 MNm^{-2}, 600°C
1150°C	–	–	970	1338	25·0	124	6·0	3731
1150°C	WARM WORK	–	1020	1370	24·0	85	16·8	148,000 d
1150°C	HOT WORK	–	968	1360	18·0	98	7·6	3000
< 1140°C	–	1050°C, ·09 cm/s	1057	1386	24·1			46,960 d
< 1140°C	–	1050°C, 1·27 cm/s	1089	1389	20·5			60,900 d
< 1140°C	–	1100°C, ·09 cm/s	1045	1381	23·2			57,200 d
< 1140°C	–	1100°C, 1·27 cm/s	1103	1394	24·1			58,300 d
> 1140°C	–	1050°C, ·09 cm/s	1051	1387	15·6			3343
> 1140°C	–	1050°C, 13·3 cm/s	1101	1397	15·1			20,407
> 1140°C	–	1100°C, ·09 cm/s	1033	1390	10·6			5263
> 1140°C	–	1100°C, 12·3 cm/s	1087	1396	19·6			57,097

HEAT TREATMENT 4 hrs 1080°C + OQ/24 hrs 650°C + AC/8 hrs 760°C + AC. d = DISCONTINUED γ' SOLVUS = 1160°C

TABLE 3 PROPERTIES OF APK1-T/P

H.I.P. TREATMENT	CONVENTIONAL FORGE	ISOTHERMAL FORGE	650°C TENSILE RESULTS		
			·2% P.S. (MN/m^2)	U.T.S. (MN/m^2)	ELONGATION (%)
< 1140°C	–	–	1080	1376	13·0
< 1140°C	1100°C, 38 cm/s	–	1132	1394	17·4
< 1140°C	–	980°C, ·09 cm/s	1097	1390	14·3

HEAT TREATMENT AS FOR TABLE 3.

responsible for the good mechanical properties, particularly low cycle fatigue strength.

The isothermally forged material, however, indicates that this explanation is not quite complete. Differences in properties occur when the material is HIPed above and below the γ' solvus temperature (about 1140^{o}c for this alloy). Necklace structures occur only in the above solvus HIPed samples, the below solvus being a uniform fine grain size. In this case the necklace structures only give good low cycle fatigue properties when the forging rate is relatively high (i.e. when the strain rate is approaching that experienced in conventional forging).

Quite reasonable properties (with the exception of low cycle fatigue) are obtained by a simple HIP followed by a normal heat treatment. This suggests that production direct to sonnic shape is possible with its attendant reduction in production costs. T/P powder showed good properties both in the forged and simply HIPed condition and although the low cycle fatigue results have not been given, it is stated that they are intermediate between those of normal powder HIPed and heat treated and normal powder HIPed and hot forged. In general the mechanical properties obtained in all cases were superior to those obtained in convention cast and wrought materials such as Nimonic 901, Inconel 718 and Waspaloy.

The metallographic structures obtained by the various treatments appear to be complex, ranging from fully recrystallized fine grains to necklace structures. Their dependence on the many possible manufacturing variations are only poorly understood and more work will be rquired in this area.

5.3 IN 100 Powder

Work on the consolidation of IN 100 powder has been conducted by a number of European laboratories. Consolidation methods have included HIP, forging and extrusion and combinations of these. The observed mechanical properties are summarized in Table 4.

Workers at M.T.U.[24] have looked at HIP and forge of IN 100 powder with a relatively high oxygen content ($\sim$ 150 ppm). HIPing was conducted either below or above the γ' solvus and no prior particle boundaries were noted, probably because of a reduction in carbon content of the alloy. Table 4 shows that the tensile properties were generally lower the higher the HIP temperature but in general properties were at least as good as conventional cast and forged disc materials.

Work at ONERA and SNECMA[20,21] used HIP treatments as well as simple extrusion and a rapid isostatic pressing technique (RHIP)[30]. This last technique involves compacting material in an extrusion press with a blind die and it differs from conventional HIP in that much higher pressures are used for short times. The compacted billets were subsequently heat treated, in one instance above the γ^1 solvus (at 1220°C) and in another below (at 1120°C). The as worked compacts showed differences in properties but these differences were removed after heat treatment and virtually no differences remained. The low cycle fatigue properties were considerably better than cast INCO 718.

A small number of tests have been carried out at Fiat[23] on IN 100 which illustrate the effects of HIP either below or above the γ' solvus. Although the strength properties are not greatly affected, the ductility of high temperature HIPed material is improved.

TABLE 4 PROPERTIES OF IN 100

REF.	H.I.P. TREATMENT	CONVENTIONAL FORGE	EXTRUSION	HEAT TREAT.	650°C TENSILE			L.C.F. CYCLES
					·2% P.S. MNm^{-2}	U.T.S. MNm^{-2}	ELONG %	1050 MNm^{-2}, 600°C
24	1100°C	1160°C	-	-	970	1200	-	4000
	1160°C	1160°C	-	-	920	1130	-	3000
21, 28	-	-	1150°C	-	† 1150	1650	30	-
	-	-	1150°C	-	1150	1630	31	-
	UNKNOWN TEMP	-	-	-	1070	1450	32	-
	RHIP	-	-	-	890	1300	40	-
	-	-	1150°C	1220°C				-
	UNKNOWN TEMP	-	-	1220°C	930	1200	31	-
	RHIP	-	-	1220°C				-
	-	-	1150°C	1120°C				-
	UNKNOWN TEMP	-	-	1120°C	1090	1400	33	-
	RHIP	-	-	1120°C				-
23	1200°C	-	-	-	1050	1208	12	-
	1160°C	-	-	-	940	1150	6	-
3,4	1190°C ϕ	-	-	-	942	1286	21	-
	1230°C ϕ	-	-	-	987	1307	14·3	-
	1190°C	-	-	-	929	1050	1·9	-
	1230°C	-	-	-	1028	1028	0·2	-
	1190°C ϕ	1140°C	-	-	1000	1200	8	-
	1190°C ϕ		1140°C	-	1000	1350	18	-

† - ALL AT 550°C   ϕ - REP POWDER

Creusot-Loire [3,4] have examined the effects of HIP treatment temperature and also the effects of subsequent extrusion and forging. Some differences were observed between CLET and AA powders, the atomized powders generally having very poor ductilities. This was probably due to the higher oxygen contents of the powders. HIPed material showed a decrease in ductility with increasing HIP temperature, but all the temperatures were above γ' solvus. The HIP and extrude and HIP and forge specimens showed less ductility than the straightforward HIPed components.

5.4 U 700 and U 710 Powders

Considerable work on U 700 has been carried out at MTU [24,32]. In the initial report [21] gas atomized U 700 was consolidated by extrusion plus forging, HIP, and HIP plus conventional forging. The authors found that HIP temperature had no noticeable effect over the temperature range 1050°C to 1300°C. For extruded and forged material properties somewhat superior to conventional cast and wrought were obtained but materials which were HIPed only generally showed poorer properties. HIPed and forged materials also showed strengths which were improved and even low cycle fatigue strengths were above conventionally produced Waspalloy. Later work [32] has emphasized that M.T.U. are more interested in cost reduction by saving metal use than improving properties relative to existing alloys. Single stage hot isostatic pressing at temperatures between 1110°C and 1180°C have produced discs of net dimensions with 650°C strength, creep resistance and low cycle fatigue strength considerably better than Waspalloy. Depending on the geometry of the component involved, cost reductions of up to 50% have been claimed, but no detailed figures have been quoted.

Creusot-Loire[4] have looked at HIPed and forged U 710. HIPing took place at 1150°C and 1180°C and subsequent forging at 1140°C. The properties achieved were essentially indistinguishable from those of IN 100 studied in the same investigation. The authors conclude that, although there is some room for improvement in ductility values, the powder route materials have properties which are at least comparable with conventionally produced alloys and that they further present a considerable economic advantage over the latter.

5.5 Astroloy Powders

Work by ONERA on powder Astroloy has been reported in two publications [21,28] and reports on HIPed astroloy have been published from Fiat[23,32]. The ONERA work included consolidation by extrusion, HIP and RHIP. The Italian work suggests that there is no effect of HIP temperature on either strength or ductility at 650°C, the low ductility associated with HIPing IN 100 at 1100°C being absent. The work at ONERA suggests from the mechanical properties that Astroloy retains the memory of different consolidation procedures even after heat treatment, presumably because of the very pronounced carbide decoration of the prior particle boundaries. However, the tensile properties were found alwyas to be superior to cast and wrought material with the possible exception of high temperature ductility.

Some interesting work on the microstructure of powder metallurgy Astroloy have been reported by Jurbert and Strudel[33]. Powder Astroloy was consolidated by either HIP at 1120°C or RHIP at 1180°C or 1120°C, the conditions being chosen to ensure full densification. Following pressing specimens were annealed at various temperatures above and below the γ' solvus temperature (1140°C). During consolidation at 1180°C, there was evidence in electron micrographs of dynamic recrystallization and considerable areas were recrystallized. However, below 1140°C, recrystallization was confined to grain boundary regions only (necklace structures). Annealing of these worked materials at 1080°C had different effects. The 1120°C consolidated alloy gave fine scale recrystallization, but there was little change in the 1180°C consolidated metal. Annealing above the γ' solvus temperature tended to destroy any fine scale substructure that had been developed. The mechanical properties of these structures are not yet available, but they will be of very considerable interest in helping to evaluate the structure/properties relationships in these alloys.

5.6 General Comments

In view of the number of different alloys and different production routes which have been used for disc materials, it is difficult to draw general conclusions. However, it is clear that for all the alloys discussed and for nearly all production routes, there is no problem in producing adequate strength levels. Figure 2 shows the UTS values of the powder disc materials reported as a function of temperature and also typical values of cast and wrought Waspaloy and Astroloy. However, there is some doubt about ductility (Figure 3) and the best way to optimize low cycle fatigue properties is not at all evident. Thus ductility appears to be affected by HIP temperature, but different investigators have noticed low ductilities for both below γ' solvus and above γ' solvus treatments. Similar effects have been observed for low cycle fatigue. There is a general belief that necklace structures are good for fatigue resistance, but this is clearly not always so and, further, there is some conflict as to the best means of attaining such structures. However, the routes which appear to have considerable promise as routine production schedules are HIP to sonnic shape or HIP plus isothermal forge.

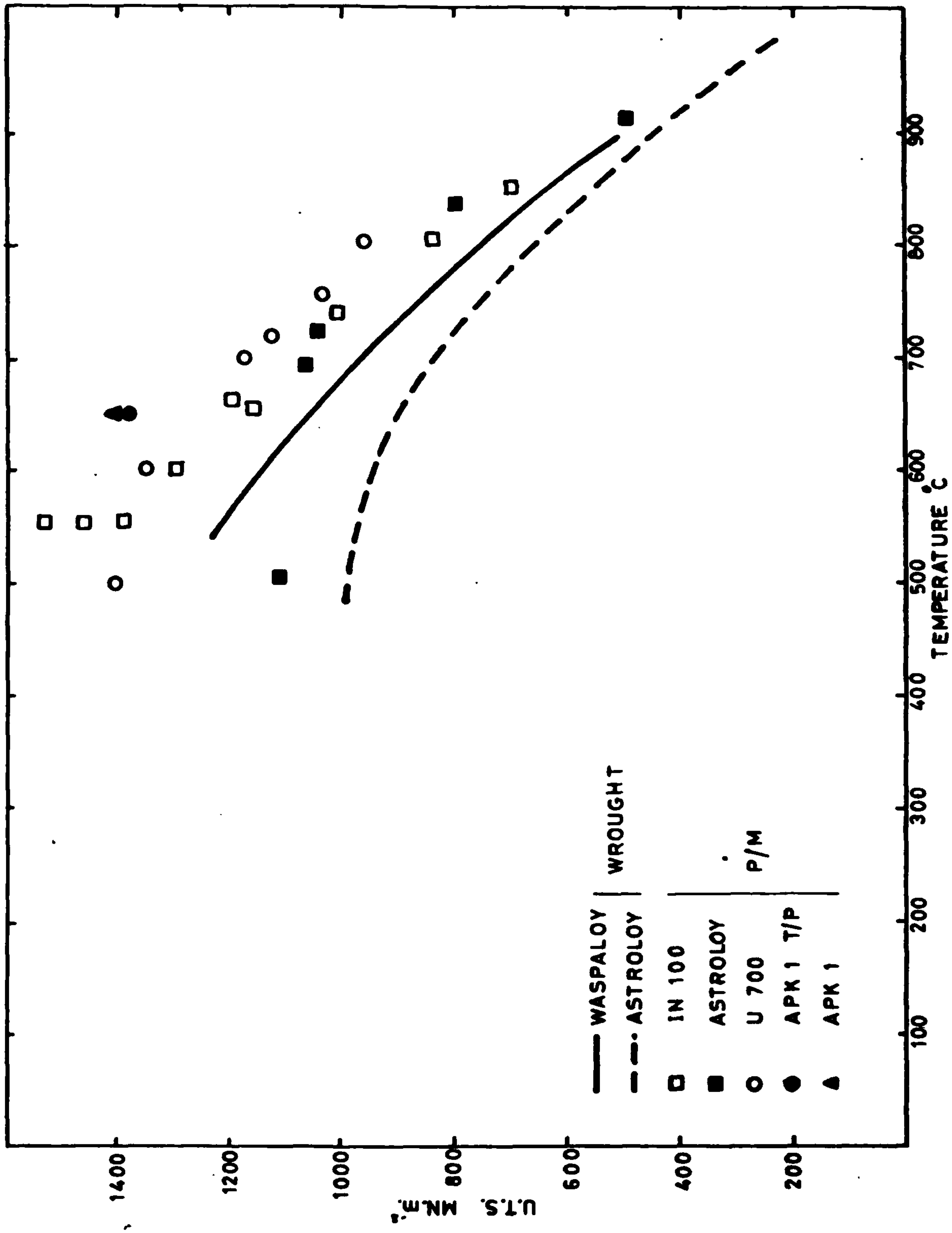


FIG 2 U.T.S. OF P/M DISC ALLOYS

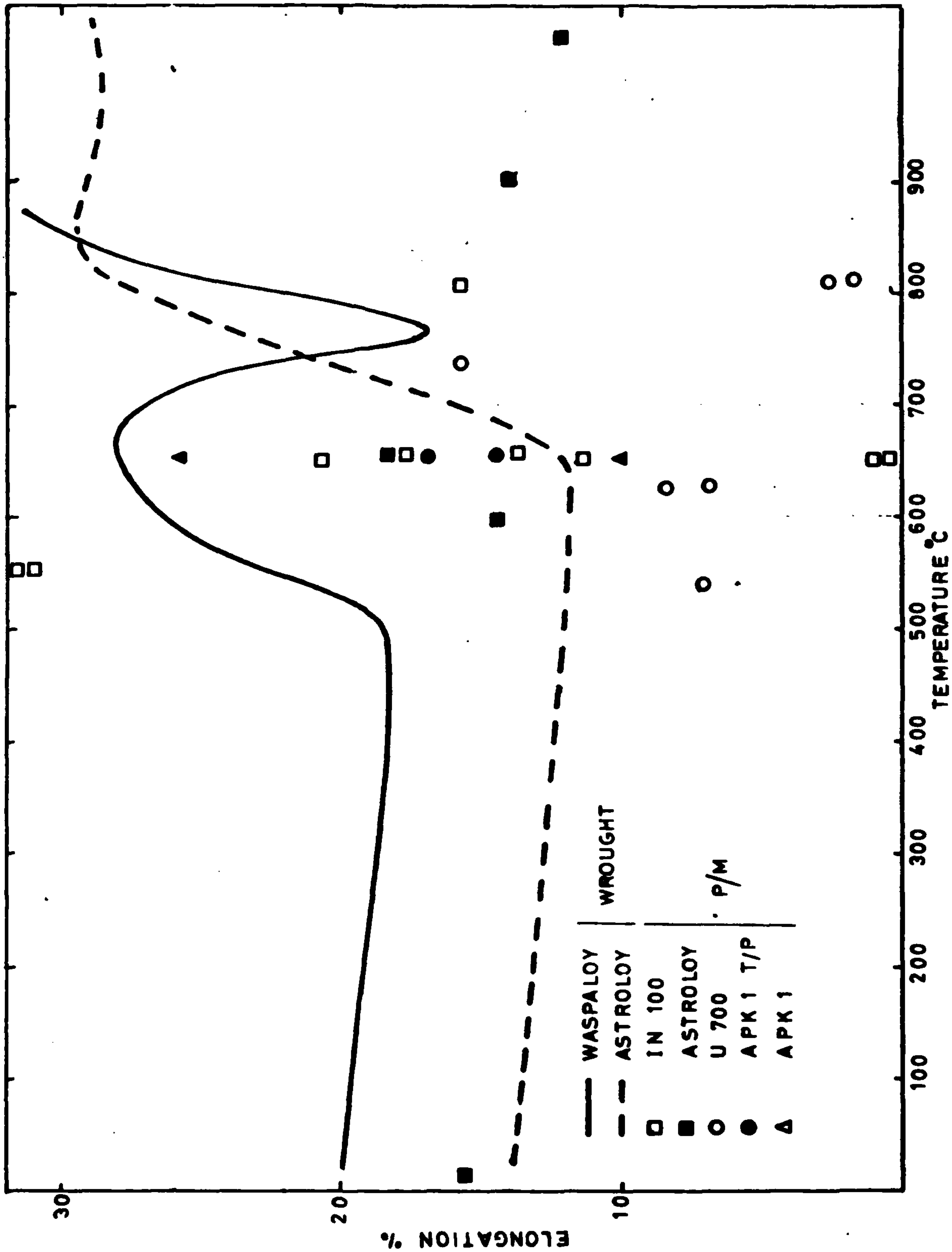


FIG 3 DUCTILITY OF P M DISC ALLOYS

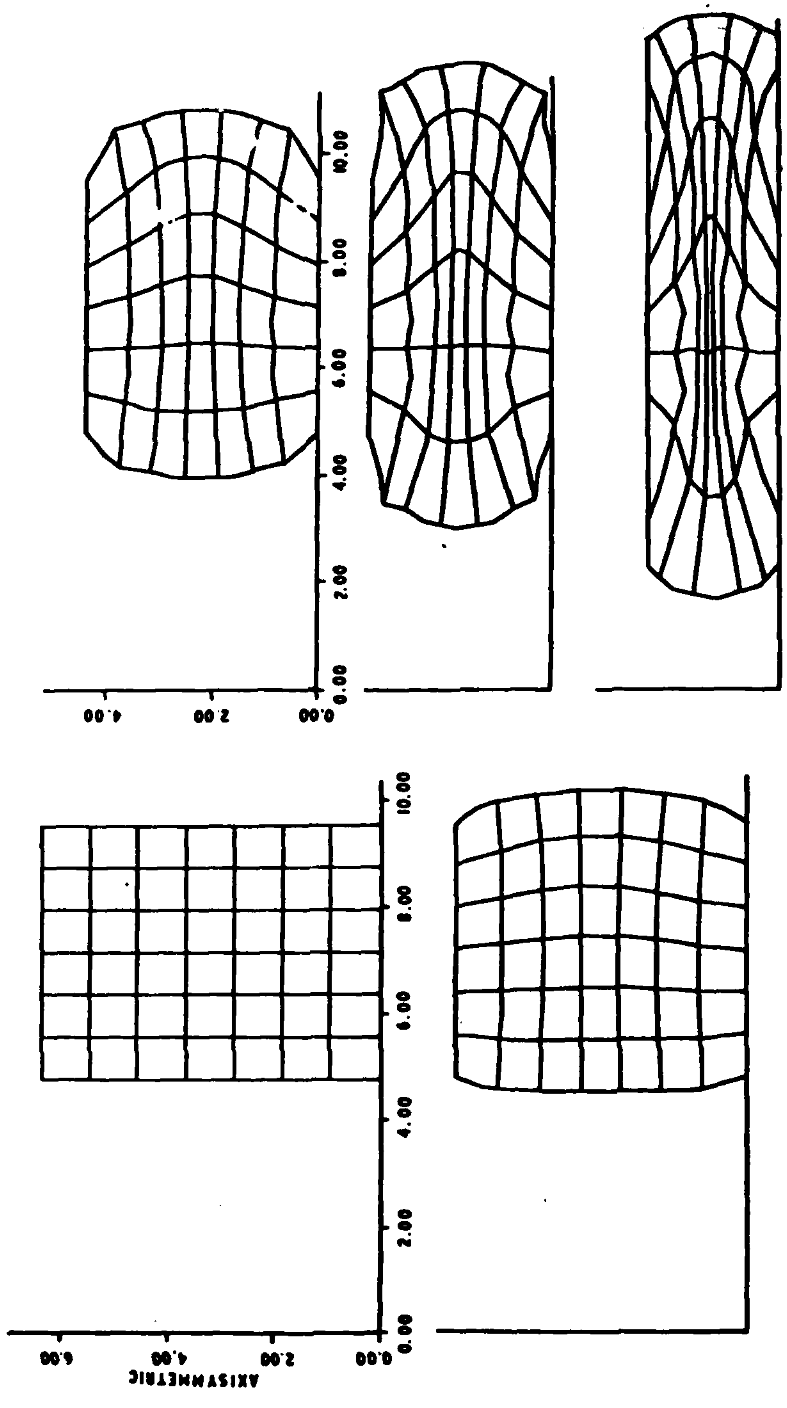
Research work is in progress in several laboratories in an attempt to clarify further the relationship between forging variables and structure/properties relationships[34,35] for APK1. Attempts are being made to draw 'structure maps' relating structure to strain, strain rate and temperature during the forging operation and to pre and post heat treatments. It is intended that such correlations should shed light on some of the apparent inconsistencies in disc properties. However, the problem is a formidable one. The occurrence of dynamic recrystallization (and hence 'warm worked' and 'necklace' structures) is a complex function of specimen structure, strain, strain rate and temperature[36,37], but it is known that these structures have an important effect on hot ductility[38,39]. All these factors must be taken into account. In addition, no hot working operation is ever homogeneous with respect to strain and strain rate[40] and as these quantities vary from place to place in a disc forging, different structures and hence different mechanical properties might be expected to develop.

A theoretical model capable of predicting strain and strain rate inhomogeneities in forged discs is vital before progress can be made. A number of possible theoretical approaches are possible[40]. Upper bound methods have been applied to a number of processes including powder forging[40,41] but they suffer from the disadvantages that the deformation pattern must be known in advance and that their sub-division of the deforming part is relatively crude. Slip line field methods are inappropriate since they assume totally unrealistic material properties (essentially rigid-plastic material). The technique which suffers from none of these disadvantages and is capable of predicting strain patterns and taking into account complex boundary conditions, complex shapes and complex material properties is the finite element analysis.

Recent applications to forging of metals have been based on the Visco-plastic approach[43,44,45,46,47] and a typical example of flow patterns is shown in Fig. 4. Work at the University College of Swansea[48] has developed a finite element procedure which will also allow the investigation of conventional forging where die cooling and adiabatic heating are allowed to affect material properties. The model can also allow for work hardening and softening and predicts not only flow patterns but the stress set up in various parts of the forging. This technique is at present being applied to help in understanding the forging/structure relationship in Nimonic 108 and AIK1[48]. It is evident that these techniques will be a powerful tool in helping to understand structure development and in future, to design forging preforms with optimum shapes for the production of good mechanical properties throughout forged discs.

5.7 Disc Forging Limitations

In spite of the good metallurgical progress made towards economical disc production from pre-alloyed powder, the maximum economy of HIP to net shape is unlikely to be achieved because of problems associated with non destructive testing[49]. The testing techniques at present require rectilinear shapes and hence considerable machining is still required after testing. More refined testing techniques are being developed but these are unlikely to be able to detect defects smaller than the equivalent of 0.015" flat bottom hole. If use is to be made of the improved strength of some powder alloy discs, then calculations of maximum allowable defect sizes show the detection limits will have to be as small as 0.010". This may well be a difficult obstacle to overcome and may prevent full advantage being taken of the stronger powder discs.



Ring test with sticking friction, half a specimen is shown. Final compression is 38.8%

FIG 4 VISCO-PLASTIC FINITE ELEMENT ANALYSIS

6. Nickel-Based Blade Materials

Two general approaches have been made towards the problem of producing blade material from powder sources. Creep strength is the critical parameter and one method of achieving the necessary levels at high temperatures is to use hardening particles which do not go into solution (as does γ'). This has led to the development of oxide and carbide dispersion strengthened alloys. These materials will be dealt separately in a subsequent section. The second approach has been to attempt to develop suitable microstructure in conventional powder metallurgy alloys. The major problems are that the powder metals are intrinsically fine grained (andhence have poor creep resistance) and that in order to reduce the influence of prior particle boundary precipitation they are often low carbon as well. Considerable attention has been paid to the problem of increasing grain size in order to improve creep resistance.

One method has been used by Claussen, Linck and Rüuchle of the Max-Planck-Institute and Daimler Benz[50]. Coarse powder of Nimonic 80A were produced by exiting the detachment of liquid drops from a refractory nozzle by induced vertical oscillations. Powder having diameters between 1 mm and 2 mm were produced with dendrite spacings of the order of 15 μm. These powders were consolidated by hot extrusion in stainless steel cans. High annealing temperatures were required to give large grains, probably because of the inhibiting effect of interdendritic carbides. Cyclic annealing above and below the TiC solvus gave a further improvement in grain growth. No difficulties were reported with hot forging of this material and although no mechanical properties results have been reported, the feasibility of the method has been reasonably demonstrated.

A different method of producing grain growth has been investigated by workers at the Chalmers University of Technology. Attempts have been made to coarsen grain size by cold working to a critical strain and subsequent annealing. Work has been conducted on a number of alloys including Astroloy[51,52,53,54], U 700[51,52] and IN 738[51 52]. Powder consolidation was accomplished by HIP and hot extrusion and in all cases, grain sizes of the order of 0.2 mm could be produced by critical cold strains of the order of 1% - 2% and a recrystallization anneal. This strain was greater than that required for nucleation, but the extra stored energy allowed migrating boundaries to pass prior particle boundaries with ease. Mechanical property testing confirmed that the coarse grained material had superior creep resistance at 950°C to the normal extruded stock. At the same time, the increase in grain size did not appear to reduce low cycle fatigue strength to any extent. Although at an early stage, this grain coarsening process holds some hope for the manufacture of powder metallurgy turbine blades.

7. Cobalt Alloys

Cobalt powder alloys have been considered for disc use at about 650°C. Work has been reported from C.R.M.[5,55,56] and ONERA[57]. A range of experimental alloys has been used and they have been designed to use solid solution strengthening or hardening by either carbides or precipitation of γ' Co_3Ti. Marty and Walder[57] concentrated on chromium carbide hardening and produced a range of alloys whose carbon contents varied but were always chosen to be either eutectoid or hypoeutectoid in composition. Some of the alloys were of a simple CoCrC type but others contained about 11% Ni in order to stabilize the high temperature face centered cubic phase. The powders were prepared by rotating electrode methods and consolidation was by hot extrusion at either 1050°C or 1100°C. All the alloys showed fully recrystallized grains with diameters between 2 and 5 μm and an even dispersion of chromium carbide whose size increased somewhat with increasing extrusion temperature. The tensile properties of the alloys at 550°C are summarized in Table 5. Even at this relatively low temperature, the strengths of the alloys are inferior to those of nickel-based material and the ductilities are very much worse.

In addition to carbide strengthening, workers at ONERA have investigated alloys strengthened by Co_3Ti. The powders of all alloys were extruded at temperatures between 1050°C and 1200°C and then further hot worked by swaging or rolling at 1100°C. The mechanical properties at 650°C are shown in Table 5. Carbide strengthening (alloys V0 to V5) is only moderately effective. Greater strengths are shown by the Titanium containing alloys, but these high strengths are associated with severe ductility problems, especially for fine grained material. The authors report great difficulty in eliminating prior particle boundaries in the titanium strengthened alloys and it is clear that powders with very low oxygen and carbon contents

TABLE 5 COBALT ALLOYS

REFERENCE	ALLOY	EXTRUSION TREATMENT	TENSILE PROPERTIES TEMP °C	·2% P.S. MN m^{-2}	U.T.S MN m^{-2}	ELONG. %	L.C.F. AT 550°C 750 MN m^{-2}
57	CoCrC	1050/1100°C	550°C	1300	1450	1	
	CoNiCrC	1050/1100°C	550°C	900	1100	1.5	
	19 KCN	1050/1100°C	550°C	920	1120	3	
	15 KCN	1050/1100°C	550°C	850	1080	7	
	13 KCN	1050/1100°C	550°C	750	1000	10.5	
	X 40	1050/1100°C	550°C	650	1200	14.0	4000 CYCLES †
5, 55	V0	1060/1200°C	650°C		1010	12	
	V1	1050/1200°C	650°C		990	12	
	V2	1060/1200°C	650°C		950	13.5	
	V3	1050/1200°C	650°C		820	15	
	V4	1050/1200°C	650°C		810	9	
	V5	1050/1200°C	650°C		800	7	
	V6	1050/1200°C	650°C		1150	9	
	V13	1050/1200°C	650°C		1200	19	
	V14 ϕ	1050/1200°C	650°C		1350	4/14	
	V14 *	1050/1200°C	650°C		1120	8/15	

† CORRESPONDING VALUE FOR IN 100 = 16,000 CYCLES. ϕ - FINE GRAINED * - COARSE GRAINED

will be required to overcome this problem. Grain coarsening could eventually only be achieved by giving large hot reductions by rolling at 1100^{o}C (above the solvus temperature for the prior particle boundary intermetallics).

The creep properties of the carbide strengthened materials are considerably better than the wrought commercial X40. The test temperature was 650^{o}C where considerable grain boundary sliding occurs and the presence of a large number of grain boundary carbides has improved creep resistance. It is possible that the combination of carbide strengthening with Co_3Ti strengthening may give an overall improvement in properties but there is the difficulty that the titanium added will react with the carbon present to form stable carbides and hence de-stabilize the γ'. In general there appear to be, at present, no great advantages in the use of cobalt base alloys as disc materials since their strengths are never better than nickel alloys and their ductilities are generally much worse.

8. Titanium Alloys

The α-titanium alloys, which have a moderate high temperature capability have been investigated at CENG[13] and by Hillnhagen and Kramer at Krupp, Essen[58]. In both cases the alloy Ti-6Al-5Zr was used. Other investigations[13,19,22] have examined the (α + β) alloy Ti-6Al-4V.

Vaughn et al.[22] investigated both REP and CSC powders of Ti-6Al-4V. Powders were HIPed at 950°C followed by a hot roll at 950°C and a heat treatment consisting of anneal at 960°C, air cool and 2 hours at 700°C. The room temperature tensile properties were indistinguishable from wrought material both with regard to strength and ductility. Low cycle fatigue at room temperature showed a very considerable scatter and in general the fatigue resistance of the powders was poorer than cast material. The authors noticed that fatigue failure nearly always started at inclusions which were thought to have come from the non-consumable tungsten electrode. Very similar results have been reported by Keinath[19] who HIPed powders at temperatures between 800°C and 950°C. Reasonable room temperature tensile properties were again observed but fatigue strength was low. The fatigue cracks were associated with inclusions. Workers at CENG consolidated their EBRE powders by hot extrusion at temperatures between 850°C and 1050°C. They observed room temperature tensile properties which were as good as wrought products but no fatigue results were quoted.

The object of the work on Ti-6Al-5Zr was to investigate the usefulness of the powder alloy for use in engine components up to about 500°C. Powders were prepared by EBRE at CENG and consolidation was carried out by HIP at a range of temperatures in both the α + β ranges. A sequence of tensile tests were carried out at various temperatures. At room temperature both the strength and ductility were better than wrought material but at high temperatures (∿ 400°C) they were somewhat poorer. Low cycle fatigue tests at room temperature and above showed a similar variation. The decrease in low cycle fatigue strength was particularly marked when powders which had been contaminated by small amounts of previously atomized copper, iron and aluminium were used.

It is clear that purity of titanium alloy powders is critical in achieving adequate fatigue properties. This has considerable implications for production methods and emphasizes the importance of designing equipment which allows 'good housekeeping' to be achieved with relative ease.

9. Dispersion Strengthened Alloys

Oxide dispersion strengthened superalloys for blade applications have been investigated by Brown Boveri, Krupp, Bofors and Sulzer. Interest has centered on Y_2O_3 strengthened IN 738 and projects have investigated mechanical alloying, compaction, hot workability as well as mechanical properties and oxidation resistance.

The alloy preparation technique was, in all cases, dry ball milling[59,60]. Reactive elements were added as master alloys and large particles were used in order to decrease oxygen contamination and give a more homogeneous composite powder. The mechanically alloyed powders[59] were in a state of supersaturated solid solution and high stored energy of deformation. Two methods of consolidation were investigated. Brown Boveri proceeded by extrusion[61] at temperatures between 950°C and 1150°C and at extrusion ratios between 9:1 and 20:1. The hardness of extruded material decreased with increasing temperature up to 1050°C but further rises in temperature to above the γ' solvus temperature had no further effect. Subsequent annealing at 1250°C gave a drop in hardness, indicating that appreciable stored energy of deformation had been retained from the extrusion process. HIP consolidation was investigated by Krupp[59,62]. Cans were HIPed at 950°C and 1050°C. The extruded and HIPed material all showed a recrystallized grain structure, the grain size increasing with increasing consolidation temperature and were in the range 0.2 μm to 1.4 μm, the smaller sizes being associated with extrusion.

For good creep properties a large elongated grain size is required[63] and grain coarsening experiments were performed. Annealing of extruded material gave a highly anisotropic grain growth, grain growth rates being very rapid in the longitudinal direction of the rods. Annealing at 1270°C gave a longitudinal grain diameter of 500 μm with a transverse diameter of

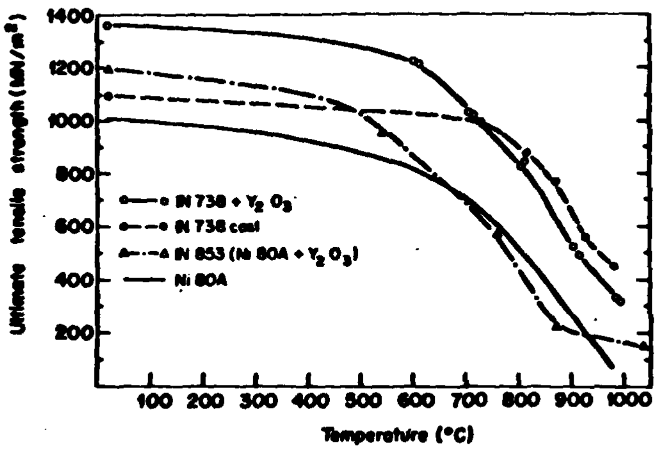


FIG 5 TENSILE STRENGTH OF IN-738 + Y_2O_3

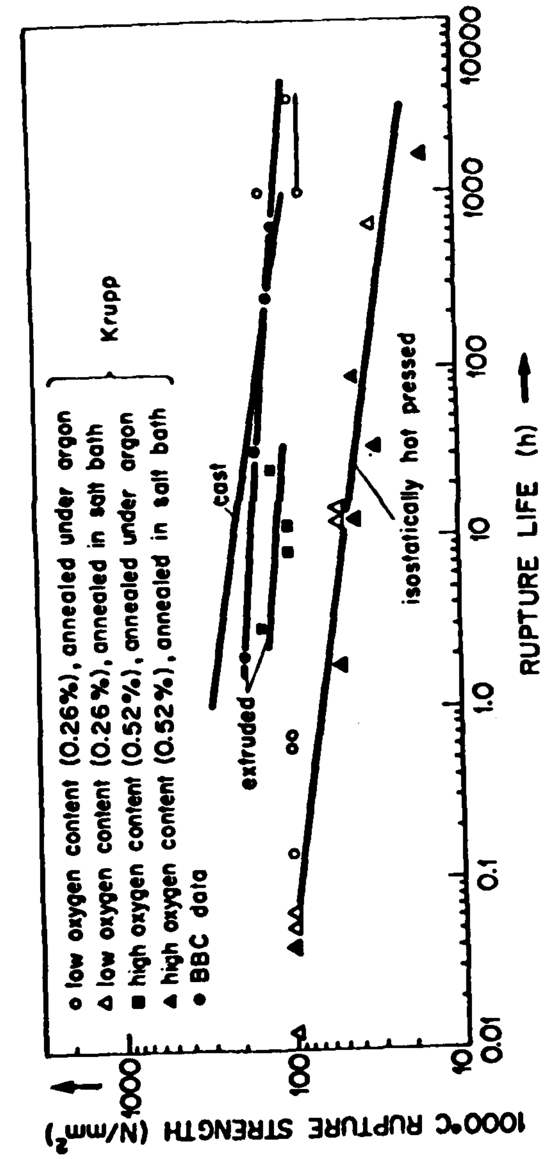


FIG 6 CREEP RUPTURE OF IN 738 + Y_2O_3

about 100 μm. Grain growth was very rapid(64,65) and the development of large grains depended on achieving a high heating rate. Since very rapid heating rates gave local transient melting and hence Y_2O_3 coarsening, an optimum heating rate of 50s to 1270°C was used. Grain coarsening was much more difficult in HIPed material and no elongated grains could be produced. After careful heat treatment equiaxed grains of about 500 μm were obtained.

The dispersion hardened IN 738 requires further fabrication to blade shapes, and an investigation of forging behaviour was carried out by Bofors (66,67). The inherently fine grained extruded material had a low flow stress and high ductility at forging temperatures but it was vital that any forging retained the ability to undergo massive grain growth on annealing. The best conditions were found to be a forge at 1100°C with a strain rate of 2.5 s^{-1}. Under these conditions the alloy had a flow stress of 60 MN m^{-2} and a ductility of 70%. It will clearly be necessary to use an isothermal forging procedure to attain these values in practice.

The mechanical properties of the alloys were measured by high temperature tensile testing and by creep testing at 1000°C. The U.T.S. values are shown in Fig. 5 together with data for other alloys and cast materials. At temperatures up to 720°C, IN 738-Y_2O_3 is stronger than cast material but at higher temperatures is slightly weaker. The creep data at 1000°C is shown in Fig. 6. The extruded alloys show a higher creep strnegth and a flatter slope on the stress/rupture time graph than cast material and this suggests that the material will be useful for long life service environments. The HIPed specimens are always much weaker than the cast alloy and there appears to be a detrimental affect of oxygen on properties.

Oxidation tests[68] on the dispersion strengthened alloys with various grain sizes were carried out at temperatures between 850°C and 950°C. At all temperatures, the dispersion strengthened alloy had better oxidation resistance than in the cast. In conclusion, the results on IN 738 - Y_2O_3 are encouraging and it should be possible to improve their properties further. For instance, some further attention to achieving large grain aspect ratios should lead to an even flatter stress/time to rupture graph since such slopes have been observed with other oxide dispersion strengthened alloys[63].

The oxide strengthened blade alloy IN 738 is still only at a early stage of development, but the oxide dispersion strengthened ferritic sheet alloy IN 956E[69] is available commercially. The high temperature corrosion resistance of this alloy is good and could therefore be used uncoated for turbine vanes and hot section parts in turbines. Some work at Brown Boveri[70] has investigated the recrystallization of this alloy with a view to improving properties. The original paper describes[69] an annealing treatment which is just isothermal annealing but it is shown[70] that if directional recrystallization is carried out the grain aspect ratio can be improved from the original 6-8 to about 80. This should give rise to an appreciable improvement in creep properties, but no results have been published.

A more fundamental search for nickel base alloys strengthened by oxides and carbides and fabricated by powder metallurgy has been conducted by Fleming[71,72,73,74]. Al_2O_3 and TiC and TaC were added to elemental nickel powder and Ni/Cr powders and fabricated by cold isostatic pressing, sintering at 1300°C and, in some cases, by impact extrusion. Some high temperature properties were measured and the creep strength was found to be dependent on the volume fraction of carbides. At high temperature, strength decreased due to Ostwald ripening of carbides. Some mechanisms of strengthening were discussed, but the research did not attempt to evaluate the commercial usefulness of the alloys.

Fig. 7 shows a comparison of the creep properties of IN 738 - Y_2O_3 with other very high strength blade materials including the advanced ODS alloys MA 755E, Alloy D and two directionally solidified alloys (γ/γ' eutectic and DS MAR-M 200 + HF). Although the data is limited, IN 738 has similar properties to these alloys even though its present grain size is considerably less than the 3000 μm normally associated with cast alloys.

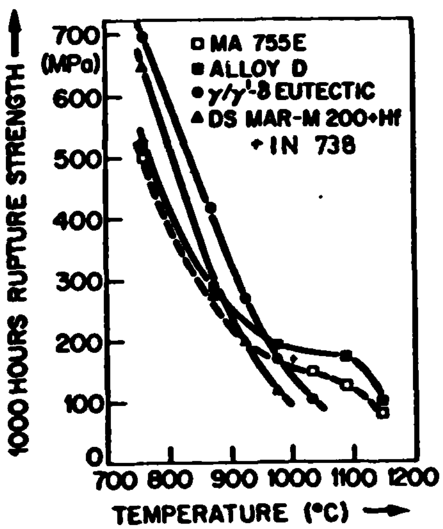


FIG 7 CREEP RUPTURE OF IN 738 + Y_2O_3 AND OTHER BLADE ALLOYS

10. Metal Spraying Techniques

One of the major motivations for powder metallurgy superalloy engine parts is that the method of production reduces chemical inhomogeniety. The segregation often encountered in highly alloyed billets is reduced to dimensions less than a single particle and the rapid cooling rates usually ensure that the dendrite size is usually much smaller than this. An alternative route by which similar micro scale segregation can be achieved in the spray forming method.

In this procedure[75] liquid metal is atomized either by inert-gas or centrifugally but no attempt is made to solidify particles and collect them as powder. Instead, the still liquid particles are made to impinge on a cold substrate and flatten and solidify in situ. The subsequent impingement of further liquid droplets can build up a deposit which is ready for subsequent hot rolling or forging. The liquid particles generally have diameters of about 100 μ and splat to pancakes of thickness ∿ 20 μ and diameter ∿ 500 μ. The cooling rates are between 10^4 and 10^6 °C/sec so that the dendrite size is quite small (∿ 5 μm). Depending on the spray density, various types of deposit can be formed. At high spray densities, new droplets impact before old droplets have completely solidified and large columnar grains develop which cross particle boundaries. At low spray densities the individual crystals are confined to within splats. Deposits of density approximately 0.98% theoretical can be formed with some discrete porosity.

The process is being explored in a number of ways. Spray rolling has been investigated at Swansea[76,77,78] and uniform strips of thickness 5 mm have been produced, but not in superalloys. Centrifugal spray forming[79] uses centrifugal atomization and Nimonic 80A has been successfully spray cast with room temperature strengths of 800 MNm^{-2} and ductilities of 28%. This material was in the form of 400 mm diameter tubes. Spray casting of ingots is also under investigation[80].

The 'Osprey' process sprays gas atomized metal into a shaped mould in order to produce a preform for subsequent forging[81,82,83]. It is claimed that considerable cost savings are possible over conventional powder metallurgy routes. A Ni-20Cr-6Mo-20Co-2Ti alloy has been successfully forged to give a tensile strength of 741 MNm^{-2} and a ductility of 28% at 780°C.

Although the processes have interesting possibilities including the possible addition of dispersed oxide strengtheners during the spraying process, they are still at an early development stage and many questions remain unanswered. For spray casting there appear to be difficulties with preform weight monitoring and in all processes, overspraying raise problems. Very little information on properties, particularly fatigue properties, is available. This is particularly important since the spray deposition takes place in argon and entrapped argon levels may be high (the porosity remaining after spray is known to be disconnected). Even quite low argon levels[1] are known to give severe problems with thermally induced porosity during heat treatment and subsequent poor fatigue resistance. This aspect clearly needs much more investigation. From this point of view, centrifugal spray casting may be the most useful since this can be carried out in vacuum.

11. Conclusions

1. Argon Atomizing is capable of producing nickel-based superalloys of good purity with high yields and throughput. The large plant at Wiggin, U.K., is capable of supplying European needs for high quality powders for the forseeable future.

2. Titanium alloy powders cannot, at present, be produced with the required purity, necessary throughput and at an economic price. No one process is demonstratably superior to the others and lack of good powder is likely to hinder research into powder metallurgy alloy properties.

3. Nickel-based powder disc materials have improved forgability. This is particularly true of T/P material and this route shows promise provided that the problem of interstitial pick up can be overcome.

4. Powder Metallurgy disc nickel alloys have been produced by a number of consolidation procedures. The most useful appears to be H.I.P., but most processes give strengths which are superior to wrought alloys.

5. The situation with respect to ductility and low cycle fatigue resistance is less satisfactory. Considerable scatter has been reported and some ductilities have been very poor. This lack of reproducibility appears to stem from a lack of basic understanding of the relationship between thermoplastic working, structure and properties. Intensive research work is being carried out in this area.

6. The field of working/structure/properties relationships is a difficult one. Little progress will be made until a thorough understanding of the mechanics of the consolidation processes is available. The most promising avenue for this work is through the finite element method.

7. Full advantage cannot be taken of the higher strengths of powder metallurgy disc materials because of the limitations of non-destructive testing techniques.

8. Few figures for materials savings in the near net shape approach are available. However, the limitations of non destructive testing again mean that the maximum materials savings cannot be made. Discs are limited to sonnic rather near net shape.

9. In spite of the considerable development and research work still required, some powder metallurgy discs are at present undergoing engine trials in the United Kingdom.

10. No new composition nickel based alloys for the powder metallurgy route are being seriously researched.

11. Powder Metallurgy Cobalt alloys appear to offer no advantages over nickel-based materials.

12. Work on blade materials is still at the laboratory stage. Preliminary results for Y_2O_3 strengthened IN 738 suggest good creep properties and the microstructures obtained suggest that further improvements in properties will be available with the system provided that heat treatments can be optimized.

13. Work on titanium alloys is being hampered by powder production. Much remains to be done on the properties of alloys since preliminary results indicate that inclusions and impurities have very deleterious effects on fatigue properties.

14. The new technique of spray forming appears to offer considerable advantages. It appears to combine the benefits of powder metallurgy routes with those of rapidly solidified alloys and to eliminate the powder handling stages. The various spray processes deserve considerable research attention in the near future.

APPENDIX 1 ALLOY COMPOSITIONS (WT %)

NAME	Ni	Cr	Co	Mo	W	Ta	Nb	Al	Ti	Fe	C	B	Zr	V	OTHERS
ASTROLOY	BAL.	15.0	17.0	5.3	-	-	-	4.0	3.5	-	0.06	0.03	-	-	-
U-700	BAL.	15.0	18.5	5.2	-	-	-	4.3	3.5	-	0.08	0.03	-	-	-
U-710	BAL.	13.0	15.0	3.0	1.5	-	-	2.5	5.0	0.5	0.07	0.02	-	-	-
WASPALOY	BAL.	19.5	13.5	4.3	-	-	-	1.3	3.0	-	0.08	0.006	0.06	-	-
APK-1	BAL.	15.0	17.0	5.0	-	-	-	4.0	3.5	-	0.025	0.02	0.05	-	-
IN-100	BAL.	10.0	15.0	3.0	-	-	-	5.4	4.5	-	0.06	0.014	0.06	-	-
IN-738	BAL.	16.0	8.5	1.7	2.6	1.7	0.9	3.4	3.4	-	0.17	0.01	0.10	-	-
MA 956 E	-	19.1	-	-	-	-	-	4.6	0.5	BAL.	-	-	-	-	0.5 Y_2O_3
Ti 6AL 4V	-	-	-	-	-	-	-	6.5	BAL.	-	0.065	-	-	4.2	-
Ti 6AL 5Zr	-	-	-	0.56	-	-	-	6.3	BAL.	-	0.023	-	5.0	-	0.15 Si
X40	10.5	25.5	BAL.	10.0	7.5	-	-	-	-	-	0.50	-	-	-	-
V0	14.9	20.1	BAL	10.5	-	-	-	-	-	-	0.03	-	-	-	-
V1	11.0	20.1	BAL	10.2	-	-	-	-	-	-	0.29	-	-	-	-
V2	14.7	20.1	BAL	10.4	-	-	-	-	-	-	0.65	-	-	-	-
V3	14.5	20.1	BAL	10.0	-	-	-	-	-	-	0.95	-	-	-	-
V4	14.7	20.0	BAL	10.0	-	-	-	-	-	-	1.36	-	-	-	-
V5	14.8	20.0	BAL	10.0	-	-	-	-	-	-	1.80	-	-	-	-
V6	-	15.7	BAL.	6.0	-	-	-	-	5.1	-	0.05	-	-	-	-
V13	-	15.3	BAL.	3.3	-	-	-	-	4.5	-	0.130	-	-	-	-
V14	-	16.2	BAL.	3.0	-	-	-	-	5.0	-	0.005	-	-	-	-
CoCrC	-	40.7	BAL.	-	-	-	-	-	-	-	2.40	-	-	-	-
CoNiCrC	19.6	41.3	BAL	-	-	-	-	-	-	-	1.89	-	-	-	-
19 KCN	11.6	38.0	BAL.	-	-	-	-	-	-	-	1.94	-	-	-	-
15 KCN	12.4	36.0	BAL.	-	-	-	-	-	-	-	1.54	-	-	-	-
13 KCN	11.4	31.0	BAL.	-	-	-	-	-	-	-	1.29	-	-	-	-

Appendix 2

European Organizations

Name	References
Atomic Energy Research Establishment Harwell, Oxon. U.K.	10,11,12
Bofors, Steel Laboratory, 20 Bofors, Sweden.	66,67
Brown Boveri and Co., CH-5401, Baden, Switzerland.	7,60,61,64,65,68,70
C.E.N.G., Centre d'Etudes Nucléaires de Grenoble, BP 85 Centre de Tri, 38041 Grenoble Cedex, France.	13
C.R.M., Centre de Recherches Métallurgiques, Abbaye du Val Benoit, 69, Leige, Belgium.	5,55,56
Chalmers University of Technology, Department of Engineering Technology, S-402, 20 Gottenburg, Sweden.	51,52,53,54
City of London Polytechnic, Whitechapel Road, London.	71,72,73,74
Creusot-Loire, Dépt. Etudes et Recherches des Accéries d'Imphy, 52160 Imphy, France.	3,4,8,9

Name	References
Daimler Benz, 7000 - Stuttgart, West Germany.	25,50
Ecoles des Mines de Paris, Paris, France.	33
FIAT, Direzione Progeltazione Corso Ferruii, 122 I-10141, Torino, Italy.	23,32
Henry Wiggin and Co. Ltd., Holmer Road, Hereford. HR4 9SL. U.K.	1,2,6,27,28,69
Imperial College of Science and Technology, Exhibition Road, London, U.K.	41,45,46,47
Imperial Metal Industries (Kynoch) Ltd., New Metals Division, Birmingham, B6 7BA. U.K.	11,22
Krupp, F., GmbH, Krupp Forchung Institut, Mücchenerstrasse 100, 4300 Essen, West Germany.	58,59
Leybold-Heraeus GmbH, Wilhelm-Rohn-Str. 25, 6450 Hanau/Main, West Germany.	14,15
Max-Plank-Institut für Metallforschung, Stuttgart, West Germany.	25,50
Messerschmitt-Bölkow-Blohm, Ottobrun, West Germany.	19

Name	References
Motoren und Turbinen-Union, M.T.U., 8000 München 50, West Germany.	24,31
ONERA, Office Nationale d'Etudes et de Recherches Aérospatiales, 92320 Châtillon, France.	21,28,57
Osprey Metals Ltd., Neath, Wales, U.K.	81,82,83
Rolls Royce Ltd., Derby, U.K.	49
SMECMA, Société National d'Etude et Construction de Moteurs d'Aviation, Usine d'Evry-Corbeil, BP 56, 91101 Corbeil, France.	21,30
Thyssen Edelstahlwerke AG, 4150 Krefeld, West Germany.	24
University College of Swansea, Metallurgy and Materials Technology Department, Singleton Park, Swansea, U.K.	38,34,43,44,75-80

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